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TEMPER EMBRITTLEMENT IN 4140 SEAMLESS TUBING

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In November 1976, several 155mm Howitzer M185 split rings failed during various stages of manufacture. A failure analysis consisting of metallography, SEM, and mechanical testing, was undertaken. This investigation concluded that the heat treatment was responsible for embrittling the steel, thereby causing the failures. Furthermore, it was shown that the required hardness could not be achieved without seriously compromising the toughness of this material.

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INTRODUCTION

In early November 1976, several 155mm M185 split rings (Fig. 1) fabricated from SAE 4140 steel seamless tubing cracked during various stages of their manufacture. In all reported failures, the rings had been heat treated and "kinked", an operation whereby the component is plastically deformed and acquires a considerable residual tensile stress. This stress is necessary for the subsequent operation of the component.

Preliminary investigations revealed that the rings were of high hardness (R_c 49-51 versus R_c 43-48 specified) and had fractured in a brittle manner. The heat treatment specified a temper of 750°F (399°C) which is within the known temper embrittlement range for this alloy. SEM fractography revealed the primary mode of fracture to be intergranular (Fig. 2). This examination also revealed that the fracture likely initiated at a stringer-type non-metallic inclusion which intersected the metal surface. Accordingly, a metallographic inspection of a mounted specimen taken from the ring showed the microstructure to contain numerous stringers of non-metallic inclusions as well as globules of a second phase material (most likely carbide precipitates) in the grain boundaries. Surprisingly though, these grain boundary precipitates were not observed on the fracture surface during the scanning electron microscopy (SEM) examination.

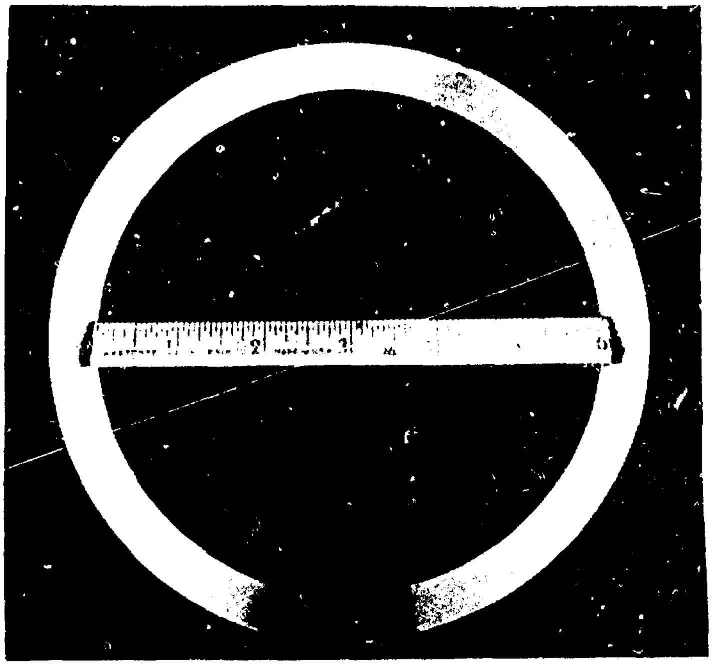

Figure 1. Overall view of split ring: Segment removed for analysis.

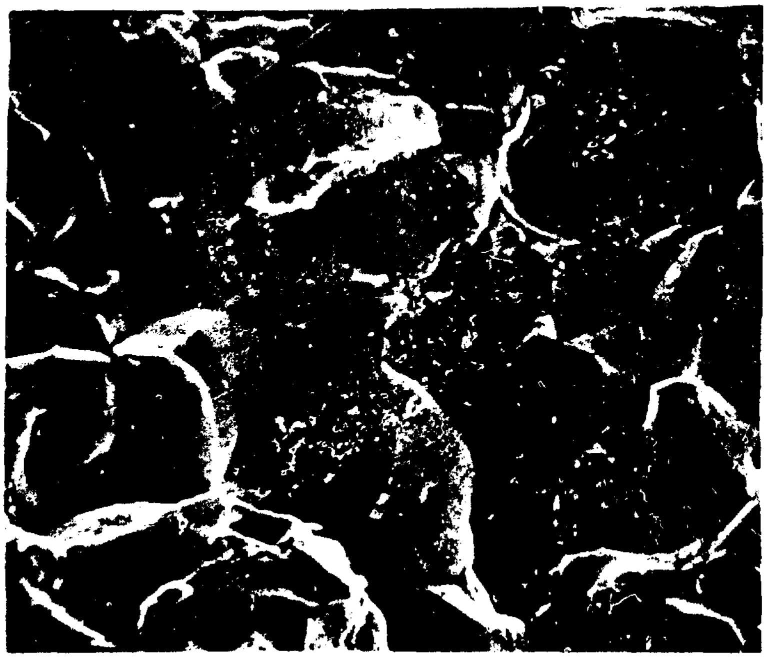

Figure 2. SEM fractograph of cracked split ring showing intergranular mode of fracture.

Nevertheless, temper embrittlement was suspected, as this phenomenon manifests itself as a loss of toughness in a steel and a tendency to fracture along prior austenitic grain boundaries[1]. In some cases, this embrittlement is viewed to be a manifestation (the result of a buildup) of impurities (such as Sb, Sn, a, P or Se) along grain boundaries as a result of the rejection of these elements from carbide precipitates in the boundaries[2]. These characteristics of temper embrittlement (which parallel the previously mentioned conditions found in the steel) coupled with the fact that the components were tempered within a known embrittling range for SAE 4140 steel strongly suggested that temper embrittlement was indeed responsible for the failures.

An experiment was undertaken to fully assess the effects of tempering on the mechanical properties of this material and determine if the current thermal treatment was responsible for embrittling the steel. This work, accompanied by metallographic and SEM fractography of the fracture surface of test specimens established the susceptibility of SAE 4140 seamless tubing to embrittlement at various tempering temperatures.

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1. McMahon, C.J., "Temper Embrittlement of Steel", ASM, 1964, p. 127

2. Rellick, J.R. and McMahon, J., "Intergranular Embrittlement of Iron-Carbon Alloys by Impurities", METALLURGICAL TRANSACTIONS, 1974, vol. 5, p. 2449.

PROCEDURE

A section of the 4140 seamless tubing was cut into disks for tensile and Charpy impact (CVN) test coupons. These coupons were divided into five sets, each to receive a different tempering treatment. Each set was austenitized at 1550°F (843°C) for 1-1/2 hours immediately followed by an oil quench. The specimens were tempered for two hours as follows:

Group	Tempering Temperature °F(°C)	Quench Method
I	500 (260)	Water
II	600 (316)	Water
III	750 (399)	Air
IV	750 (399)	Water
V	800 (427)	Water
VI	900 (482)	Water
VII	1000 (538)	Water

Group III was considered the control group in the experiment as it approximates the reported heat treatment that the rings receive in manufacturing.

Impact tests (according to ASTM spec. A370-72a) were conducted over a temperature range from -150°F (-101°C) to +212°F (100°C) to obtain the ductile to brittle fracture transition behavior for each tempered condition. The impact toughness data were then plotted versus testing temperature for each of the five heat treatments. These graphs allow a reasonably accurate estimate of the transition temperature (T.T.), where the steel exhibits a change in fracture mode from ductile

to brittle. Ductile failure implies a mechanism where considerable energy is absorbed while a brittle failure process absorbs relatively little energy.

RESULTS AND DISCUSSION

The results of the mechanical property tests and chemical analysis are summarized in Tables 1 and 2. As expected, the transition temperature increased with decreasing tempering temperature. In fact, the T.T. was above room temperature for tempers of 800°F (427°C) and below. This is very undesirable and it is evident that the required hardness level in this alloy will consistently result in a high T.T. As a consequence, once a crack is initiated in this material, it meets very little resistance in propagating through the component as long as an applied stress is present. In this situation, the stress is available from the kinking operation which plastically deforms the material and results in a high residual stress in the piece. The tempers of 900°F (482°C) and 1000°F (538°C) markedly improved the T.T. property but did not meet the required hardness specification of R_c 43-48. Also, the -40° Charpy impact properties for the 750°F (399°C), 800°F (427°C) tempering temperatures were poor (6-6.5 ft-lbs). Again, the 900°F (482°C) and 1000°F (538°C) tempers improved this but, as previously mentioned, did not meet the hardness specification given for the split ring components.

TABLE 1. MECHANICAL PROPERTIES

TEMPERING TEMPERATURE °F(°C)	DUCTILE TO BRITTLE TRANSITION TEMP. °F(°C)	HARDNESS Rc	YIELD STRENGTH 0.1% (ksi)	-40°F CHARPY IMPACT(ft-lb)	K_{Ic} * ($ksi\text{-}in^{1/2}$)
500 (260)	200 (93)	50.5	219.1	3.5	-
600 (316)	200 (93)	49.0	209.1	3.2	-
750 (399)	94 (34)	45.5	203.8	6.5	36.5
750 (399)	82 (28)	46.0	201.0	6.5	44.3
800 (427)	74 (23)	44.0	185.4	6.0	79.0
900 (482)	18 (-8)	40.0	176.4	12.0	99.3
1000 (538)	-72 (-58)	37.0	150.8	24.5	128.7

*Predicted fracture toughness based on correlation of Barsom and Rolfe.

TABLE 2. CHEMICAL ANALYSIS

C	Mn	Si	Cr	Mo	Sb	Sn
.38	.84	.23	.98	.23	.08	trace

An estimate of fracture toughness for the thermal treatments was determined via a correlation of Barsom and Rolfe from the "high energy" impact toughness data[3]. As shown in Table 1, the 750°F (399°C) and 800°F (427°C) tempers produce poor fracture toughness properties with systematic improvements achieved by the 900°F (482°C) and 1000°F (538°C) tempers. This indicates that the stress necessary to cause failure and/or the defect size necessary to initiate a fracture is decreasing with lower tempering temperature. As previously mentioned, this steel contained numerous stringer type non-metallics which can act in the manner of a crack or flaw. The overall result is a material whose ability to resist crack initiation and propagation is seriously impaired.

Metallographic examination revealed a martensitic microstructure in all five heats as shown in Figure 3. This examination also revealed the presence of a precipitate in the grain boundaries (Figure 4). SEM examination complemented the optical microscopy by revealing the nodular voids left by chemically removing carbides from the austenitic grain boundaries (Figure 5).

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3. Barsom, J. M., and Rolfe, S. T., "Impact Testing of Metals," STP 466, 1970.

Figure 3. Microstructure for 1000°F temper - 1000X.

Figure 4. Grain boundary precipitates - 1000X.

Figure [illegible]. SEM micrograph displaying grain boundary voids due to carbide precipitates - 750°F temper - [illegible]00X.

Fractographic examination of broken CVN bars indicates the increasing tendency towards "low energy" intergranular failure as opposed to "low energy" cleavage failure as the tempering temperature is lowered. The 1000°F (538°C) temper displays quasi-cleavage as the principal mode of failure (Figure 6). This mode, however, changes to a mixed mode of intergranular and quasi-cleavage in the 750°F (390°C) temper (Figure 7). The high energy mode of failure for both tempers was ductile fracture by microvoid coalescence (Figures 8-9).

It is important to note that the high hardness of the split rings investigated was obviously not achieved using the reported heat treatment the rings undergo. This was substantiated by conducting a series of heat treatments at lower tempering temperatures. Table 3 reveals the actual tempering temperature required to obtain R_c 50 hardness (in this steel) to be approximately 500°F (200°C) as opposed to the 750°F (399°C) temper reportedly given the split rings. The tempering response for 4140 steel is plotted in Figure 10.

The results of this investigation show 4140 steel tubing to be embrittled by the current tempering practice of 750°F - (399°C) air cool. Unfortunately, higher tempering temperatures will not yield the desired hardness level in this steel. Therefore, it is apparent that this particular lot of 4140

Figure 6. SEM fractograph showing cleavage failure mode - 1000°F temper - 960X.

Figure 7. SEM fractograph showing cleavage and intergranular failure modes - 750°F temper - 900X.

Figure 8. SEM fractograph showing ductile failure mode - 1000°F temper - 500X.

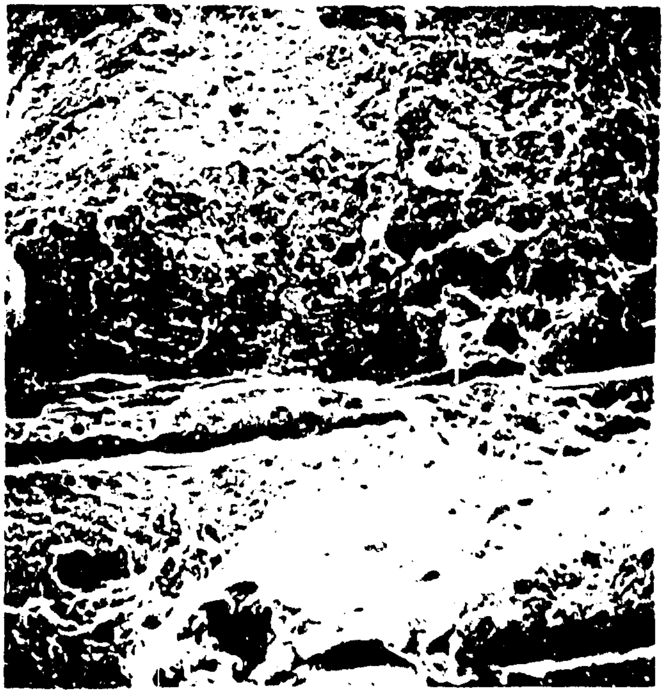

Figure 9. SEM fractograph showing ductile failure mode - 750°F temper - 500X.

TABLE 3. TEMPERING TEMPERATURE SUMMARY FOR 4140 STEEL

Tempering Temperature °F (°C)	Hardness (R_C)
1000 (538)	36
900 (482)	40
800* (427)	44
750* (399)	46
650* (343)	47.5
600 (316)	49
500 (260)	50.5
400 (204)	53
300 (149)	55
As Quenched	

*Tempers which meet hardness specification R_C 43-48

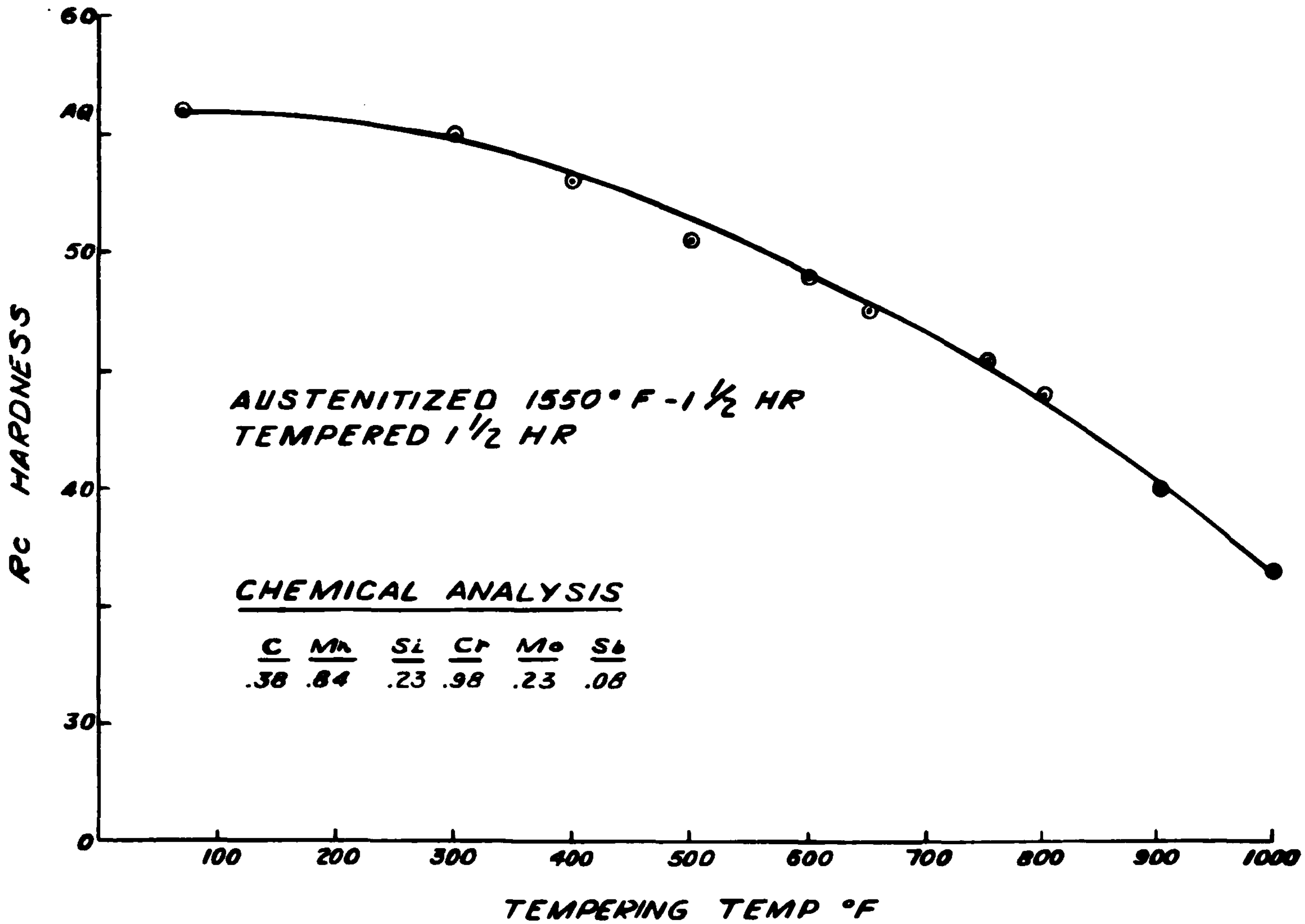

Figure 10. Tempering response of 4140 steel.

tubing will not adequately meet all the requirements imposed on the split ring. This is partially a consequence of material cleanliness (numerous, large stringers) and heat treatment. Since 4140 steel has been used successfully in this application for previous split rings, it is likely that one or more of the factors involved, viz., non-metallics, residual stress, embrittling precipitates, etc., was not present in a critical capacity previously. However, this is a probablistic situation and obviously can be easily upset by subtle changes in the variables.

CONCLUSIONS

1. SAE-4140 steel, tempered at 750°F and 800°F, exhibited relatively high ductile to brittle transition temperatures (72°F (22°C), 94°F (34°C) resp.) and low Charpy impact (-40°F) properties (6, 6.5 ft-lbs. resp.). The 1000°F and 900°F tempers improved both properties but failed to meet the split ring hardness specification.

2. The estimated fracture toughness of this steel decreases with lower tempering temperatures and is seriously degraded by the 750°F tempering treatment.

3. The microstructure was tempered martensite which contained numerous non-metallic stringers as well as a network

of carbide precipitate in the austenitic grain boundaries.

4. In the presence of high residual stress, SAE 4140 should not be used in this case unless the hardness requirement can be lowered. Otherwise, a change of alloy should be considered. Some potential candidates include; 4330 + V (Vac Arc), "Hy-Tuf" (7-550°F), 18% Ni MARAGE (200 or 250 Grade).

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